High strength corrosion-resistant steels with metastable austenite for elastic elements and springs for demanding applications

Cite as: AIP Conference Proceedings **1785**, 040036 (2016); https://doi.org/10.1063/1.4967093
Published Online: 18 November 2016

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High Strength Corrosion-Resistant Steels with Metastable Austenite for Elastic Elements and Springs for Demanding Applications

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Abstract. Obtaining a high-strength state in developed steels of austenitic and austenitic-ferritic types is investigated. The prospects of the application of these steels are demonstrated for critical elastic elements and springs to be used over a wide range of temperatures, from cryogenic to elevated (400 °C).

INTRODUCTION

The use of corrosion-resistant steels of the 12Kh18N10T type is not always reasonable, since it does not satisfy the relaxation resistance, durability, and especially ductility and adaptability to manufacture. These properties are important for the production of the thinnest wire. The need to create such materials arises due to the expanding application of above-mentioned products. These materials must be appropriate for operating at low and elevated temperatures and have more advantageous complex of physic-mechanical, corrosion, operational and technological properties than the existing materials.

New [1, 2] carbon-free aluminum-containing steels with a Fe-Cr-Ni base and Mo, Ti and Co alloying have a polyhedral austenitic structure with a large number of annealing twins in the hardened state (0.5–1.0 % Al) or an austenitic-ferritic structure (2.0–2.5 % Al), depending on aluminum content.

The investigated steels have a uniform base of alloying with low-carbon, less than 0.03 %; it is one of the preconditions for high plasticity and adaptability to manufacture of the investigated steels in the quenched state. Another factor that determines their high plasticity is the presence of strain-metastable austenite, which becomes strain-induced martensite, partially or completely. Since the amount of austenite is different in the investigated steels (≈100 % in the 03Kh14N11K5M2YuT austenitic steel, 50–30 % in the 03Kh14N10K5M2Yu2T austenitic-ferritic steel), the formation of strain-induced martensite and the TRIP effect are unequal.

The powerful TRIP effect enables one to perform cold plastic deformation with extremely high reduction ratios in the austenitic steel 03Kh14N11K5M2YuT and to obtain a nanocrystalline structure. The application of these large plastic deformations (without significant loss of ductility) significantly reduces the number of intermediate annealings in the production of thin-section wires compared to the 12Kh18N10T steel. This causes a decrease in its costs, despite the more complex composition of the new steels. As a result of cold plastic deformation by drawing, a more than 4 times increase in the strength properties compared to the quenched state is observed.

Mechanics, Resource and Diagnostics of Materials and Structures (MRDMS-2016)
AIP Conf. Proc. 1785, 040036-1–040036-4; doi: 10.1063/1.4967093
Published by AIP Publishing. 978-0-7354-1447-1/$30.00

EXPERIMENTAL

It was shown in [3] that the 03Kh14N11K5M2YuT quenched austenitic stainless steel has high plasticity and toughness and low values of strength and elastic properties: σ_u=540 MPa; $\sigma_{0.2}$=245 MПa; $\sigma_{0.03}$=230 MPa; δ=63 %; ψ=84 %; KCV=2.29 MJ/m^2. Hardened 03Kh14N10K5M2Yu2T austenitic-ferritic steel consists of 2 phases, namely, δ-ferrite and reversed austenite, ~ 50:50 ratio. In [4] it was reported that δ-ferrite has abnormally high hardness (≥ 500 HV), while the hardness of austenite is low (≈ 200 HV). The superstructure reflections of the δ-ferrite areas are prohibited for δ-ferrite in electron diffraction. The lattice type B2 is the best description of this superstructure. A dark-field image of the (001) superstructure reflections indicate the presence ("glow") round-shaped fine crystals of uniformly distributed δ–ferrite particles. The particles of the (Fe,Ni)Al intermetallic phase have a coherent relationship with BCC δ–ferrite. Despite the fact that the microhardness of δ-ferrite is high, the integral hardness and strength of this steel is low. This allows for a large total amount of strain. The mechanical properties of the hardened 03Kh14N10K5M2Yu2T austenitic-ferritic steel are as follows: σ_u=940 MPa; $\sigma_{0.2}$=780 MPa; $\sigma_{0.03}$=545 MPa; δ=55 %; ψ=71 %; KCV=1.94 MJ/m^2.

The mechanical properties of Duplex steels after cold plastic deformation by drawing are 2.5 times as high as those in the hardened state. The existing phase (γ and δ-ferrite) have different properties in the hardened austenitic-ferritic steel. As shown by the electron-microstructure examination, under any form of loading (compression, drawing with small degrees of compression, shear, deformation begins in the austenitic phase with the occurrence of local microtwinning with the subsequent formation of a dislocation cell structure and then a fragmented structure and strain-induced martensite. It is found that δ-ferrite does not undergo any deformations due to the presence of finely dispersed precipitates of the uniformly distributed ordered (Fe,Ni)Al intermetallic phase owing to its high hardness from small to moderate amounts of strain (40 %). The interphase boundaries are clear and straight.

A further increase in the degree of deformation leads to a distortion of the interphase boundaries, blurring, increased dislocation density and δ-ferrite fragmentation. Thus, 80–90 % amount of cold plastic deformation leads to the formation of 100 % of the BCC phase (50 % crushed δ-ferrite and formed crystals of strain-induced martensite).

Metastable austenite transforms into strain-induced martensite causing essential hardening of the investigated steels. After deformation, the BCC phase (100 %) is thermally stable up to 500 °C. The transformations occurring in cold samples under heating were studied by thermal X-ray analysis. Thermal X-rays showed only the BCC phase (strain-induced martensite and cold-hardened δ-ferrite in the austenitic-ferritic steel and ~ 100 % strain-induced martensite in the austenitic steel) in the structure at room temperature. As the temperature increases above 550 °C, the reverse transformation of strain-induced martensite into austenite (reversed) occurs, the phase ratio is restored (BCC and FCC 50:50), which was originally present in the structure-hardened steel. Therefore, the recommended temperature of aging is 500 °C, the soaking period being 1 hour.

Post-deformation aging, which is usually performed on the final product, provides a significant additional increase in the strength properties. The disintegration of the supersaturated BCC-solid solution (strain-induced martensite) occurs with the formation of (Fe,Ni)Al aluminide during aging, it is ordered isomorphic to the BCC phase. Transmission electron microscopy demonstrates that the precipitation of the (Fe,Ni)Al intermetallic phase is nanocrystalline size (6..10 nm). This is extremely important in the production of the thinnest wire, since the particle size in the hardening phase must be much smaller than the diameter of the wire.

RESULTS AND DISCUSSION

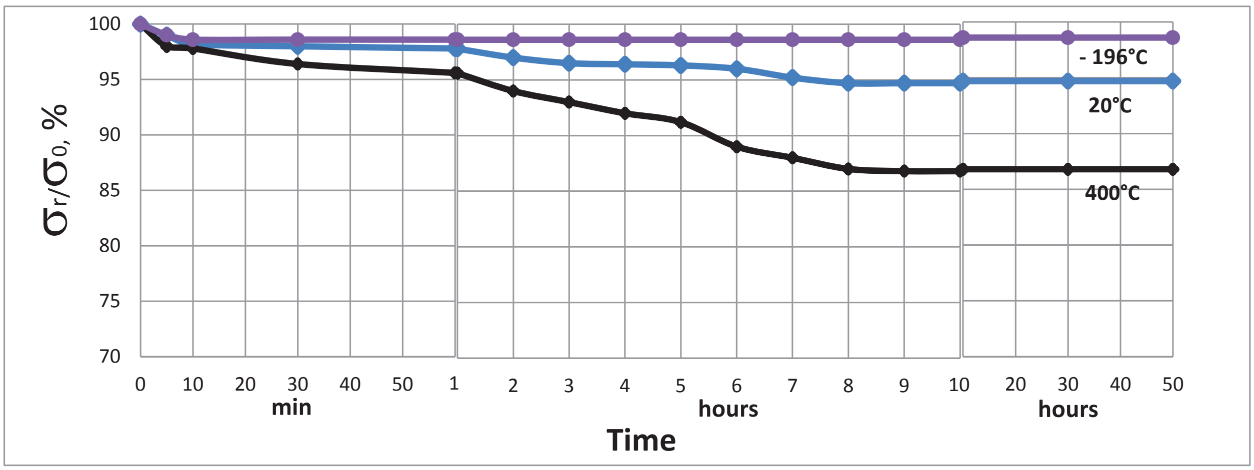
Relaxation resistance is an important characteristic at increased, room and cryogenic temperatures along with the elastic, strength and plastic properties of the spring wire and strip [5]. This must imply that the achievement of high strength is almost always based on obtaining a metastable state. The relaxation of stresses is caused by the interaction of the structural and shift mechanisms. The increase of the metastable structure during deformation and stress relaxation due to the structural mechanism is blocked by relaxation decreasing through the shear mechanism, due to the increase of the resistance to plastic deformation of the deformed investigated steels during aging. As indicated above, cold plastic deformation allows achieving a significant increase in strength and rigidity. Optional aging increases stress relaxation resistance at 400 °C, particularly, when the aging temperature becomes higher relaxation temperature, which increases the stability of the structure.

Figure 1 shows the curves of the changes in the relative relaxation resistance of the investigated steels. It was studied at 400 °C, 20 °C and cryogen temperature, trial duration being 50 hours. The samples underwent optimal treatment, which is as follows: quenching + 60% deformation + aging at 500 °C, 1h for the 03Kh14N10K5M2Yu2T austenitic steel (a) and the 03Kh14N10K5M2Yu2T austenitic-ferritic steel (b). Initial stress was 80% of the elastic limit.

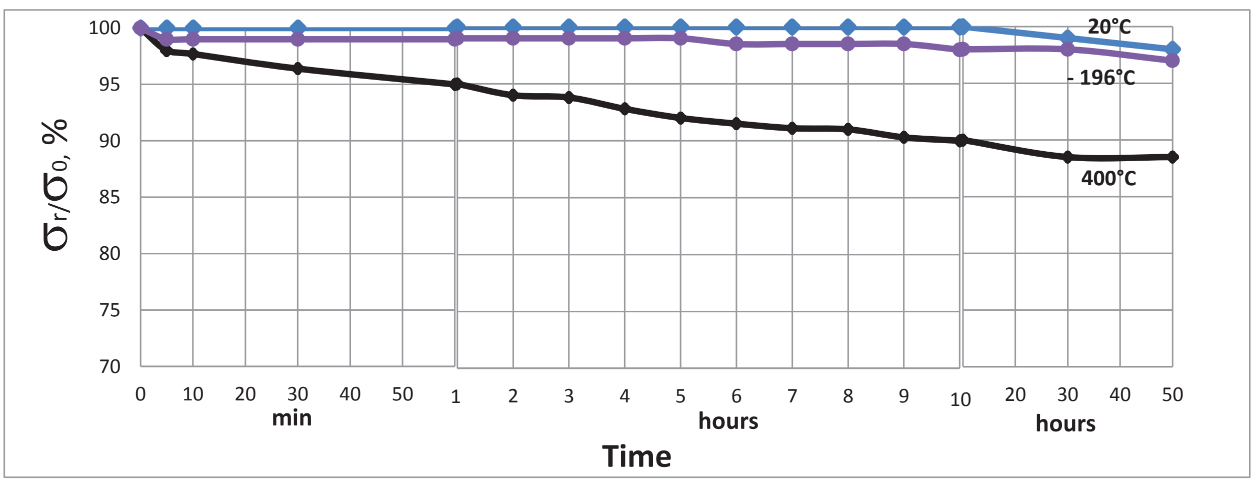
It can be concluded from these data that the relaxation resistance of the investigated steels is sufficiently high in a wide range of temperatures, from cryogenic to high (400 °C) after treatment according to the optimum mode.

The decrease of relaxing tension does not exceed 5 % at room temperature, 2 % at cryogenic, and 15 % at high temperature for the 03Kh14N10K5M2Yu2T metastable austenitic steel; 2 %, 3 % and 12 %, respectively, for the 03Kh14N10K5M2Yu2T austenitic-ferritic steel.

Magnetic characteristics were measured under deep cooling and subsequent heating to room temperature, Fig. 2. Magnetization as a function of temperature for the investigated steels was determined using a 7407 Lake Shore vibrating magnetometer in the temperature range of 77…300 K in a magnetic field ranging from 80 A/cm to 10350 A/cm. The research has shown that steel samples practically do not change their saturation magnetization after the “cooling to 77 K – heating to 300K” loop. This indicates the stability of the phase composition of these steels up to the temperature of liquid nitrogen.



(a)



(b)

FIGURE 1. Relaxation resistance of the austenitic steel (a) and the austenitic-ferritic steel (b) at room (20 °C), increased (400 °C) and cryogen temperatures

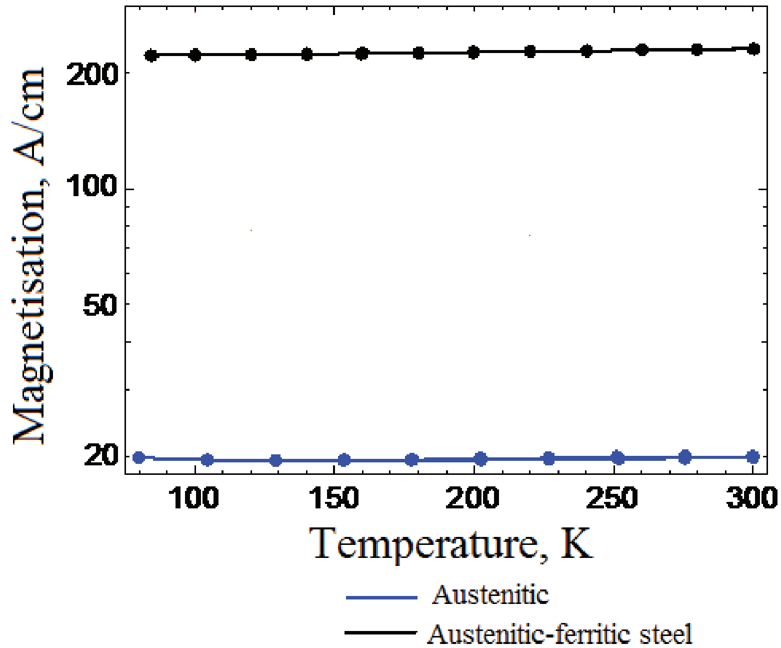


FIGURE 2. The temperature-magnetization dependence for the austenitic steel and the austenitic-ferritic steel

CONCLUSION

Thus, the investigated austenitic and austenitic-ferritic steels are promising material for the production of high-strength springs and elastic elements, devices of demanding application, which operate not only at room and high temperatures, but also at negative and cryogenic temperatures.

ACKNOWLEDGMENTS

The work was supported by the Russian Government Resolution 211, contract No. 02.A03.21.0006, research 2014/236 for the Ministry of Education and Science of the Russian Federation.

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